

Undissociated screw dislocations in silicon: Calculations of core structure and energy

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Abstract

The stability of the perfect screw dislocation in silicon has been investigated using both classical potentials and first principles calculations. Although a recent study stated that the stable screw was located both in the 'shuffle' and 'glide' sets of $\{111\}$ planes (Koizumi *et al*, 2000, Phil. Mag. A, 80, 609), it is shown that this result depends on the classical potential used, and that the most stable configuration belongs to the 'shuffle' set only, in the centre of one $(\bar{1}01)$ hexagon. We also investigated the stability of an sp^2 hybridization in the core of the dislocation, obtained for one metastable configuration in the 'glide' set. The core structures are characterized in several ways, with a description of the three dimensional structure, differential displacement maps, and derivatives of the dis-registry.

1 Introduction

Dislocations in silicon have been the subject of many investigations, both experimental, computational and theoretical, not only because they can appear in microelectronic devices, but also because of their own properties, closely related to the covalent nature of bonding in this material. In ordinary conditions, silicon is brittle below about 600°C (Hirsch, Samuels and Roberts 1989, George and Michot 1993). Above this temperature, TEM observations show dissociated dislocations (Ray and Cockayne 1971) which thus lie between the narrowly spaced $\{111\}$ planes *i.e.* belonging to the glide set. On the basis of computed generalized stacking-fault energy and entropy calculations, on $\{111\}$ narrowly spaced (glide set) and widely spaced (shuffle set) respectively (Kaxiras and Duesbery 1993), the idea was put forward of the possibility of a transition :

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at low temperature, perfect dislocations are easier to nucleate and move in the shuffle set, while the activation barrier for glide becomes lower for dissociated dislocations in the glide set at high temperature (Duesbery and Joos 1996).

Thus, experiments have been undertaken in which, in order to achieve plastic deformation at temperatures as low as room temperature, the silicon sample is either submitted to a high confining pressure, of the order of 5 GPa (Rabier *et al* 2001) or subject to a surface scratch test (Rabier *et al* 2000). Under these conditions, the microstructure has been found formed of non-dissociated dislocations, supposed to belong to the shuffle set. Favoured dislocation orientations appear to be screw, 60° , 30° and also 41° . Similarly, deformation experiments in III-V compounds such as GaAs, InP, InSb, performed down to 77K by applying a high confining pressure (Suzuki *et al* 1998, 1999a, b), indicate that the low temperature plastic deformation is governed by kink pair formation on undissociated screw dislocations moving in the shuffle set planes.

Theoretical investigations of the core structures of dislocations in silicon are then clearly required to bring additional insights. However, despite a large number of existing atomistic computations, most of these were devoted to partial dislocations of the glide set, with a particular attention to core reconstructions of the 30° and 90° partials and to mobility properties (cf. for instance the review by Bulatov *et al* 2001). Less information is available about perfect dislocations: In his pioneer examination of dislocation cores in diamond cubic structures Hornstra (1958) quite naturally placed the screw dislocation line furthest away from any atom row, that is at the center of the hexagon formed by six neighbouring $\langle 110 \rangle$ dense atom rows; the dislocation then belongs to the shuffle set, more precisely it is located at the intersection of two $\{111\}$ shuffle planes at 71.53° . By somewhat artificially rebonding atoms, Hornstra also proposed another structure for the screw core, the interesting idea being that the core is spread over two adjacent hexagons sharing a common small edge. Arias and Joannopoulos (1994) performed DFT calculations of the shuffle screw in silicon: they found the first configuration proposed by Hornstra to be stable with respect to spontaneous dissociation and they calculated energy parameters for the core. Finally, in a recent study, Koizumi, Kamimura and Suzuki (2000) have investigated the core configuration and the mobility of the $(a/2)\langle 110 \rangle$ screw using the Stillinger-Weber (1985) potential for silicon. They found two stable configurations: the configuration at the centre of the hexagon, denoted A, has a higher energy and the lower energy configuration (denoted B) can be regarded as belonging to both a $\{111\}$ shuffle plane and the $\{111\}$ glide plane at 71.53° . The authors discussed in detail the very special part that configuration B might play in cross-slip mechanism and in the transition of dislocation glide from the shuffle set to the glide set. It remains to be confirmed if these results are general and not specific to the Stillinger-Weber potential. Regarding the glide set, previous works essentially focused on partial dislocations, and, to our knowledge, there are no available studies of the perfect screw configurations.

Thus, it is of interest to investigate the core properties of perfect dislocations in the shuffle and glide sets, and particularly the existence and relative stability of all proposed configurations of the screw orientation. This paper reports on such calculations, using both (i) the Density Functional Theory (DFT) formalism and (ii) several semi-empirical potentials. After a description of the methods and computational details, we present the different energetic and structural parameters associated with each configuration. These quantities are then discussed in relation with the previous results, dislocation mobilities and validity of the classical potentials.

2 Computational methods

The atomistic calculations have first been carried out using semi-empirical potentials. We have employed (i) the SW potential of Stillinger-Weber (1985), used as reference and also for comparison with previous work from Koizumi *et al* (2000), (ii) the Tersoff potential (1988) which is able to give a better representation of a number of defects than Stillinger-Weber, and (iii) EDIP (Justo *et al* 1998) constructed so as to benefit from the successes of earlier potentials and incorporating data obtained from DFT calculations, such as the gamma-surfaces. It has to be noted that for dislocation calculations, EDIP is the only semi-empirical potential able to account for reconstructions of both 30° and 90° partials in the glide set. The main advantage of empirical potentials is their low computational cost, which allows for a fast calculation of several configurations. Potentials suffer from limitations, implicitly related to their functional form or limited fitting database, and calculated energies may prove to be relatively inaccurate, especially for configurations involving highly distorted or broken bonds, as encountered in dislocation cores. However, by using three different types of potentials, we expect to overcome this issue and obtain reliable results.

First principles calculations have also been carried out, in order to discriminate between configurations, and to calculate more precise defect energies. In addition, comparisons with empirical potential allowed us to check their reliability for dislocation core investigations. We performed DFT calculations (Hohenberg and Kohn 1964, Kohn and Sham 1965) in the local density approximation at zero temperature with the ABINIT code (ABINIT, 2002). The ionic interactions were represented by norm-conserving pseudopotentials (Trouiller and Martins 1991). We used a plane-wave basis with an energy cutoff of 10 Ry and two special k -point along the dislocation line (Monkhorst and Pack, 1976). Tests with Generalized Gradient Approximations (GGA), a higher energy cutoff or a finer k -point sampling have also been conducted for selected configurations. We found an error on the defect energy lower than 0.5% using 5 special k -points and a cutoff of 14 Ry, and about 5% when using GGA.

At first, semi-empirical potential calculations were made, for a fast exploration of several

system sizes, thus determining the size effects of the simulation slab on the results. Suitable simulation box sizes were then selected, small enough for DFT calculations to remain tractable and large enough for the computed energies to be meaningful.

3 Simulation model

Ideally, we would consider an isolated straight screw dislocation in an infinite bulk. Along the dislocation line, provided that there is no reconstruction, we have a periodic situation with a period equal to the Burgers vector $\frac{1}{2}[110]a_0$. The use of periodic boundary conditions is then the obvious suitable choice along the dislocation line. In the plane perpendicular to the dislocation line, a long-range strain field will be generated by the dislocation, and should be taken into account in the calculation. Two different methods could be employed. In the first one, no periodic boundary conditions perpendicular to the dislocation line are applied, and only one dislocation is located in the centre of the simulation box (figure 1A). The atomic positions at the boundaries are then initialized to values calculated with elasticity theory using a numerical code adapted from ANCALC (Stroh, 1958,1962) or with a more precise model (Lehto and Öberg 1998). The computational box has to be large enough to prevent a fictitious interaction between fixed boundaries and the dislocation core. In addition, atoms located at the edges of the system are not in a bulk-like environment, and defect energies can not be directly extracted from total energy calculations. Here, we used such an approach for semi-empirical potential calculations only, because the simulation box could be enlarged at will, due to the low computational cost, and also because defined individual atomic energy allows an easy determination of defect energies. Typical computational cells involved about 10000 atoms (dimensions $133 \text{ Å} \times 132 \text{ Å} \times 11.5 \text{ Å}$ for a $(40 \times 84 \times 3)$ cell). The anisotropic elastic energy per unit length of the dislocation is given by the well known formula (Hirth and Lothe 1982)

$$E = \frac{Kb^2}{4\pi} \ln \left(\frac{R}{r_0} \right)$$

The second method involves periodic boundary conditions along the directions perpendicular to the dislocation line. In that case, annoying difficulties arise owing to discontinuities at the boundaries, in particular for a single dislocation. Spurious shear strain associated with these discontinuities could then have a significant influence on the dislocation core structure and energetics. These difficulties can be smoothed by considering dipolar (figure 1B) or quadrupolar (figure 1C) arrangements of dislocations in the cell. Whether a dipole or a quadrupole should be favoured depends on the character of the dislocation: a dipole is best suited for an edge dislocation whereas a quadrupole minimizes the residual strain associated with a quadrupole of screw dislocations (Lehto and Öberg 1998). With the second choice, four dislocations should be

included in the cell, with separation distances large enough to prevent a spurious interaction, which would lead to large cell sizes. However, as suggested by Bigger *et al* (1992), the system could be divided by a factor of two by relaxing the orthogonality constraints on the periodic cell (figure 1D). In this work, the quadrupolar arrangements of dislocations (figure 1C and 1D) have been considered for both semi-empirical and first principles calculations. The semi-empirical calculations were useful for investigating easily several cell sizes and estimating the non-elastic core-core energy contributions possibly present for very small cells. We considered computational cells ranging from $(40 \times 84 \times 3)$ to $(6 \times 12 \times 3)$. For ab initio calculations, the largest $(12 \times 12 \times 1)$ cell encompasses 144 atoms, with 2 dislocations.

For a quadrupolar distribution and four dislocations in an orthogonal cell, the anisotropic elastic energy per unit length is obtained by summation of the interactions between dislocation pairs, calculated using a code adapted from ANCALC (Stroh, 1958,1962). The total energy can be split up into an interaction energy inside the cell (E^{intra}) and half the interaction energy between the quadrupole and all its periodically repeated images (E^{inter}). A reference (zero) of the elastic energy is required for determining E^{intra} , and is chosen as the elastic energy of a dislocations quadrupole whose distance along the edge is d_0 . In that case, it can be shown that the reference distance d_0 is equal to the core radius r_0 obtained for a single dislocation. In fact, if the quadrupole is extremely large such that the four dislocations can be considered as isolated, the elastic energy amounts to four times the self energy of a single dislocation. The determination of E^{inter} should require an infinite summation, which has to be dealt with care. A quick convergence may be obtained by summing the interaction energy between quadrupoles. It has to be noted that this is not the case for a dipolar arrangement, and special handling of the summation is required (Cai *et al* 2001). The derivation of the elastic energy for a quadrupolar distribution in a non-orthogonal cell (two dislocations per cell instead of four, see figure 1) is straightforward.

4 Results

The table 1 shows the elastic constants calculated with the semi-empirical potentials and first principles. These constants are used for generating the initial configurations from anisotropic elasticity theory, and extracting core energetics from relaxed systems.

In figure 2, we show a $(\bar{1}01)$ section of the cubic diamond structure, with three possible locations of the dislocation line. A corresponds to the original position at the center of one hexagon (Hornstra 1958), for a screw dislocation belonging to two shuffle planes. B was recently proposed by Koizumi *et al* (2000), at the middle of one long hexagon bond. It is interesting to point out that in this case, the dislocation is located at the crossing of both a shuffle and a

glide $\{111\}$ plane. Finally, another high symmetry location on the structure is point C, at the middle of a short hexagon bond, with the screw dislocation belonging to two glide planes. Other locations have been investigated, either inside the hexagon or at the exact position of one Si atom, but in all cases, the system relaxed to one of the three selected configurations.

The different energetic values resulting from all our calculations are reported in table 2. The energy differences show that with ab initio and all potentials but SW, A is the most stable configuration. We were able to reproduce the results of Koizumi *et al* (2001), B appearing more stable than A by using the SW potential. B is obtained as the second choice with ab initio and EDIP potential, while it seems highly unfavourable with the Tersoff potential. Another important point concerns the stability of the B configuration. Although the relaxation with semi-empirical potentials was straightforward, the B geometry has been found extremely difficult to retain within a first principles calculation, even using initial configurations relaxed with potentials as a starting point. Some or all dislocations of the quadrupole generally evolved to an A configuration, or annihilated themselves. Only in one case were we able to relax the structure. Finally, for all calculations, C is never the most stable configuration, or is even unstable with the SW potential.

Table 2 also reports the core radii obtained by matching the elastic energy with the calculated defect energy. For the empirical potential calculations and one unique dislocation in the computational cell, the core radius r_0 is determined by considering the defect energy contained in the cylinder centered on the dislocation line, with radius R and height the Burgers vector \mathbf{b} (see formula above). The factor K determined from the calculated elastic constants is used. We considered that the core radius r_0 is already well converged for $R = 60 \text{ \AA}$. For example, the core radius changes by less than 0.01 \AA for R ranging between 40 and 60 \AA , for a configuration A and the SW potential. For first principles calculations and a quadrupolar distribution of dislocations, the core radius is determined numerically by inverting the defect elastic energy. We found that the core radius determination is already very precise for a (12×12) cell, with an uncertainty about 0.01 \AA . For the smaller (6×6) cell, a 0.1 \AA deviation from converged values was obtained. Most of the values are close to about 1 \AA , the commonly used $1/4$ of the Burgers vector. Only core radii for B and C with Tersoff and A with EDIP are slightly distant. In the table 2 are also reported core energy values, often used in the literature, which are obtained with a core radius equal to the Burgers vector.

In figure 3, differential displacement maps of the relaxed A, B and C configurations of the screw dislocations, obtained from systems with one unique dislocation and relaxed with an empirical potential. For the configuration A, the distortion is uniformly distributed on the hexagon ring encircling the dislocation line. It is also clear from the picture that the displacements are identical along the two 'shuffle' planes (see figure 2). The configuration B is characterized by a maximal distortion on the two atoms on both sides of the 'shuffle' plane. Most of the constraints

are located on the hexagon rings sharing these atoms. In the case of C, the maximal distortion is also located on two atoms, but on both sides of a 'glide' plane. Average deformations are observed for all four hexagon rings around these two atoms, and equivalent displacements along the two glide planes passing through the C screw core. It is noteworthy that for each geometry, almost identical pictures have been obtained regardless of the potential considered. One exception is the C configuration with the SW potential, which relaxes to the C' configuration represented in figure 3. The initial differential displacement located on the two atoms next to the dislocation location vanished, and large distortions appeared involving more distant atoms (see figure 3). We found this configuration to be unstable with both EDIP and Tersoff, relaxing to A. In figure 4, the differential displacement map for a quadrupolar distribution of A configuration, relaxed with first principles, is represented. Even if the dislocation cores are close and interact together, it is noteworthy that the general pattern obtained for a unique dislocation remains easily recognizable in that case.

To characterize the spatial extension of the dislocation core, we have determined the width at half maximum (WHM) of the derivative of the disregistry introduced by the dislocation. The disregistry is obtained by computing the difference of displacements along $(\bar{1}01)$ for atoms on either sides of the glide $(\bar{1}1\bar{1})$ plane (Figure 5). The calculated points are fitted on the simple shape $\arctan(\frac{x}{\Delta})$ of a dislocation (Hirth and Lothe, 1982). The determination of the WHM is then straightforward. In table 3, we report all values, as well as the WHM calculated in the same way for configurations built from anisotropic elasticity theory. In contrast to the energy ordering, it appears that the WHM's do not depend on the kind of classical potential, for a given configuration. For C, the ab initio WHM is also equal to the classical one. It is also clear that the B core is wider than the A core.

5 Discussion

The A configuration has been considered as the most plausible structure for the undissociated screw dislocation in silicon. But a recent study by Koizumi *et al* (2000) concluded that the B configuration is slightly more stable than A. Our calculations indicate that A is definitely the most stable geometry for the screw, using ab initio and two classical potentials. However, with the SW potential, we were able to reproduce the result of Koizumi *et al*, which proves that the apparent stability of B over A is an artefact of the potential. With first principles calculations involving various cell sizes, the B configuration was found to be stable only in one case. It seems that the slightest deformation could lead to the relaxation from B to A. We conclude that the B configuration is weakly metastable. Additional insights are obtained from the analysis of the core. The figure 6 shows three dimensional structures of three configurations. A is characterized

by the absence of atomic rearrangements in the core, the main distortions being located on all bonds forming the hexagon ring encircling the dislocation core. On the contrary, for B, the main deformations are located on the two atoms close to the core. They are bonded together and at the same height along $(\bar{1}01)$ in the bulk. After introduction of the dislocation, the atoms have now a height difference of half the Burgers vector along $(\bar{1}01)$, and they are separated by about 2.8 Å. Weak bonds with such an interatomic distance are possible for silicon. However, each of these atoms has already a coordination 3 and would need to accommodate 2 extra bonds (see figure 6), which is unlikely to occur. From the analysis of the electronic density, it appears that in the B core, 2 rows of dangling bonds follow the dislocation line. On the one hand, these rows may explain the very low stability of this configuration with first principles methods. On the other hand, the large range of energy values illustrates the difficulty of describing dangling bonds with the classical potentials. We investigated a possible reconstruction of the B configuration along the dislocation line, for reducing the number of dangling bonds. Because the atoms involved are located on either side of the 'shuffle' plane, new bonds are formed only at the expense of breaking other bonds. Interestingly, this by-hand reconstruction relaxed to the C configuration. The WHM, and so the core extension, of B is larger than A, which may also explain why it is less stable.

We also investigated the possibility of a screw dislocation in the glide set with the C configuration. With all classical potentials and first principles method, it is found that A is more stable than C, with a large energy difference. It is then unlikely that the C core could be formed in bulk silicon. However, interesting features are associated with this structure. The examination of the geometry in the core revealed that the two atoms on either side of the dislocation line (black balls in figure 6) have a coordination 3. Before the introduction of the dislocation, these two atoms, were bonded together, with a height difference of half the Burgers vector along $(\bar{1}01)$. After, they are still bonded together but located at the same height, one bond per atom being broken. The *ab initio* interatomic distance between these two atoms is 2.16 Å, whereas distances with neighbour atoms are about 2.29 Å. The three bonds for each atom are almost co-planar, and the angle between them range from 117° to 123°. All these quantities point at an sp^2 hybridization of these two atoms, with a double bond between them, which is confirmed by the analysis of the electronic density. This possibility has already been proposed by Hornstra (1958), on the basis of geometrical arguments only. An sp^2 hybridization is not favoured in silicon, which explained the large defect energy for the C configuration. However, it would be interesting to investigate the competition between screw configurations in diamond for example, where sp^2 is favoured over sp^3 . It is worth noting that a sp^2 character is also present in recently proposed metastable structures for the 30° and 90° partial dislocations in diamond (Ewels *et al* 2001, Blumenau *et al* 2002). One interesting aspect of the C configuration is the tightness of the core. It is difficult to

compare directly A and C since they are not located in the same family of 111 planes. Nevertheless, insights could be gained by the comparison with the anisotropic elastic solution (Table 3). It is clear that for all potentials, and especially for ab initio, the relaxed A core is wider than the initial elastic configuration. The effect is even stronger for the B configuration. However, in the case of C, the core is contracted by the atomic relaxation. The narrowness of C could be partly attributed to the formation of the double bond in the core.

To our knowledge, the A configuration has already been investigated with first principles techniques in two previous studies. Arias and Joannopoulos (1994) found a core energy $E_c = 0.56 \pm 0.21$ eV.Å⁻¹ whereas Miyata and Fujiwara (2001) obtained E_c equal to 0.95 eV.Å⁻¹. These results may be compared to our ab initio core energy of 0.52 eV.Å⁻¹ (table 2), very close to the value of the former study. However, this agreement seems fortuitous, the authors using isotropic theory and a fitted $K \equiv \mu = 0.29$ eV.Å⁻³. Miyata and Fujiwara (2001) followed a similar approach, but with a fitted $K \equiv \mu = 0.48$ eV.Å⁻³, and obtained a larger core energy. Instead, here, K factors calculated within anisotropic theory and with the ab initio calculated elastic constants, very close to the experimental values (see table 2), are employed. The disagreement between the different works may be explained by the poor k-point sampling in the newer study, as well as the use of isotropic theory with a fitted K factor.

We discussed our results in relation to the screw mobility. A being the most stable geometry, possible paths from one minimum to another include saddle configurations B or C (see Figure 2). Koizumi *et al* (2000) found a Peierls stress of about 2 GPa for the non-dissociated screw dislocation, considering the path A→B→A. However, this relatively low value may be explained by the use of the SW potential, and its failure to yield the correct stability of A and B configurations. In fact, although the Peierls stress cannot be simply determined from static calculations, several insights may be obtained from the analysis of the energy differences between configurations. With SW, the energy difference between A and B is only 0.14 eV per Burgers vector. With first principles, we determined a larger energy difference of 0.32 eV per Burgers vector. It is then reasonable to assume that the Peierls stress for the path A→B→A will be much higher than 2 GPa. This is confirmed by recent ab initio calculations from Miyata and Fujiwara (2001), where the Peierls stress ranged from 22 to 30 GPa. Another possible path for dislocation cross slip would be A→C→A. However, the large calculated energy difference indicates a very large stress, and this possibility may be ruled out solely on the basis of energy considerations.

Finally, we compared the merits of each classical potentials we have used in this study. On the basis of our three investigated configurations, it appears that EDIP is better suited than SW or the Tersoff potential for this study, the stability and energy differences being close to the ab initio results. This is not completely surprising since this potential has been designed specifically to study defects (Justo 1998). The worst is maybe the SW potential, which yields the B core as

the most stable configuration. It is worth noting that although the stabilities of the different core configurations very much depend on the kind of potential, we obtained similar atomic structures in almost all cases. Consequently, for relaxing configurations prior to ab initio calculations, one could use any classical potential. But for investigating stability, it is necessary to consider several kind of potentials.

6 Conclusion

Using anisotropic elasticity theory, several semi-empirical classical potentials, and first principles calculations, we have investigated the properties of the undissociated screw dislocation in silicon. Considering previous studies and the geometry of the silicon atomic structure, three possible structures have been selected and compared. We have shown that the configuration A, where the dislocation core is located in the centre of one hexagon, in the shuffle set, is clearly more stable than the two others. In a former study by Koizumi and Kamimura (2000), another configuration, with the dislocation located in the centre of one long hexagon edge, was favoured. From our calculations, it appears that this result is explained by the use of the Stillinger Weber potential, this configuration being less stable with other classical potentials or ab initio methods. We also investigated a third solution, with the dislocation in the glide set. Despite its high defect energy, this configuration presents the interesting feature of an sp^2 hybridization of the atoms forming the core. Obviously, such a structure is worth to be studied in a material favouring sp^2 , like diamond. We also characterized the spatial extension of the cores of each structure by determining the derivative of the disregistry. A possible continuation of this work includes the determination of the mobility of the undissociated screw dislocation in silicon. The study of other dislocation orientations, such as those recently observed at low temperature (Rabier *et al*, 2000, 2001), would be another working direction.

References

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Table caption

Table 1: Experimental (Simmons and Wang 1971) and calculated elastic constants (in Mbar) for the SW Stillinger-Weber (Stillinger and Weber 1985), Tersoff (Tersoff 1989) and EDIP (Justo *et al* 1998) potentials as well as our ab initio results. The SW parameters have been rescaled in order to fit the experimental cohesive energy of 4.63 eV. For a screw dislocation in a cubic diamond structure, $K = [C_{44}(C_{11} - C_{12})/2]^{-\frac{1}{2}}$

Table 2: Calculated energetic parameters for A, B and C screw dislocations. ΔE is the energy difference, in eV per Burgers vector, with an uncertainty of 0.01 eV/Bv. r_0 is the core radius (± 0.03 Å). E_c is the core energy, obtained with a fixed core radius equal to the Burgers vector (± 0.02 eV.Å⁻¹). Note that C is not stable with the SW potential, it relaxed to a configuration with an energy 0.02 eV/Bv higher than A.

Table 3: WHM (in Å) obtained from elasticity (with experimental C_{ij}), classical potentials, and ab initio. Two values are undetermined, due to the instability of C with SW, and the difficulty to obtain B with first principles in a large cell.

	exp.	SW	Tersoff	EDIP	ab initio
B	0.99	1.083	0.978	0.99	0.99
C_{11}	1.67	1.617	1.425	1.75	1.64
C_{12}	0.65	0.816	0.754	0.62	0.66
C_{44}	0.81	0.603	0.687	0.71	0.78
C_{44}^0		1.172	1.188	1.12	1.09
K	0.64	0.49	0.48	0.63	0.62

Table 1

	ΔE (eV/Bv)		Radius r_0 (Å)			Energy E_c (eV.Å ⁻¹)		
	$E_B - E_A$	$E_C - E_A$	A	B	C	A	B	C
SW	-0.14		0.82	0.91		0.55	0.51	
Tersoff	1.08	0.54	0.78	0.35	0.52	0.55	0.84	0.70
EDIP	0.23	0.74	1.49	1.31	0.98	0.37	0.42	0.54
ab initio	0.32	0.86	1.22	1.03	0.74	0.52	0.60	0.74

Table 2

	Elastic	SW	Tersoff	EDIP	ab initio
A	2.7	3.1	3.2	3.2	3.6
B	2.6	3.9	4.0	4.0	
C	1.2		0.9	0.9	0.9

Table 3

Figure captions

Figure 1: Models for straight dislocation simulations. In A, there is no periodic boundary conditions and only one dislocation in the computational cell (dashed red line). B, C and D show periodic boundary systems with dipolar (B) or quadrupolar (C,D) distributions of dislocations.

Figure 2: Ball and stick representation of the $(\bar{1}01)$ plane of the cubic diamond structure. The three circles A, B and C mark the positions of the dislocation line. Dashed (dotted) lines show the 'shuffle' ('glide') $\{111\}$ planes.

Figure 3: Differential displacement maps of the screw dislocation in the configurations A, B, C, and C' (obtained with SW from C). The arrows are proportional to the out-of-plane $[\bar{1}01]$ shifts between neighbour atoms introduced by the dislocation. The cross marks the position of the dislocation line and the dashed line the cut plane.

Figure 4: Differential displacement maps of the screw dislocation in the configuration A, in the quadrupolar distribution shown in the figure 1. The arrows are proportional to the out-of-plane $[\bar{1}01]$ shifts between neighbour atoms introduced by the dislocation.

Figure 5: Variation of the dis-registry in the (111) plane along the $[121]$ direction for the three configurations. The solid lines are fits with the expression $f(x) = b \left[\frac{1}{\pi} \arctan \left(\frac{x}{\Delta} \right) - \frac{1}{2} \right]$. The insert graph shows the derivative of the dis-registry for A, and the definition of the WHM.

Figure 6: Ball-and-stick representation of the cubic diamond bulk (top left) and of the three screw core configurations. A six atom ring (see figure 2) is represented by dark grey sticks, in order to show the deformation due to the dislocation and the Burgers vector. Dangling bonds (for B) and sp^2 atoms (for C) are represented by black sticks and balls. The position of the dislocation line is shown by the dashed lines.

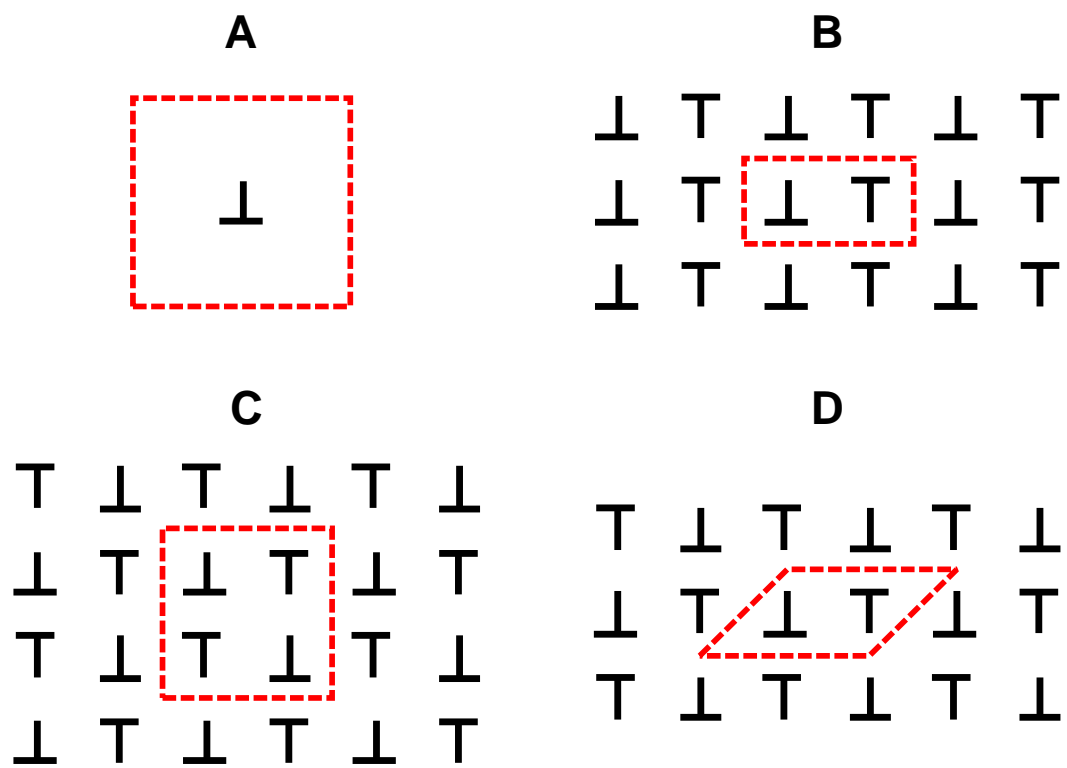


Figure 1

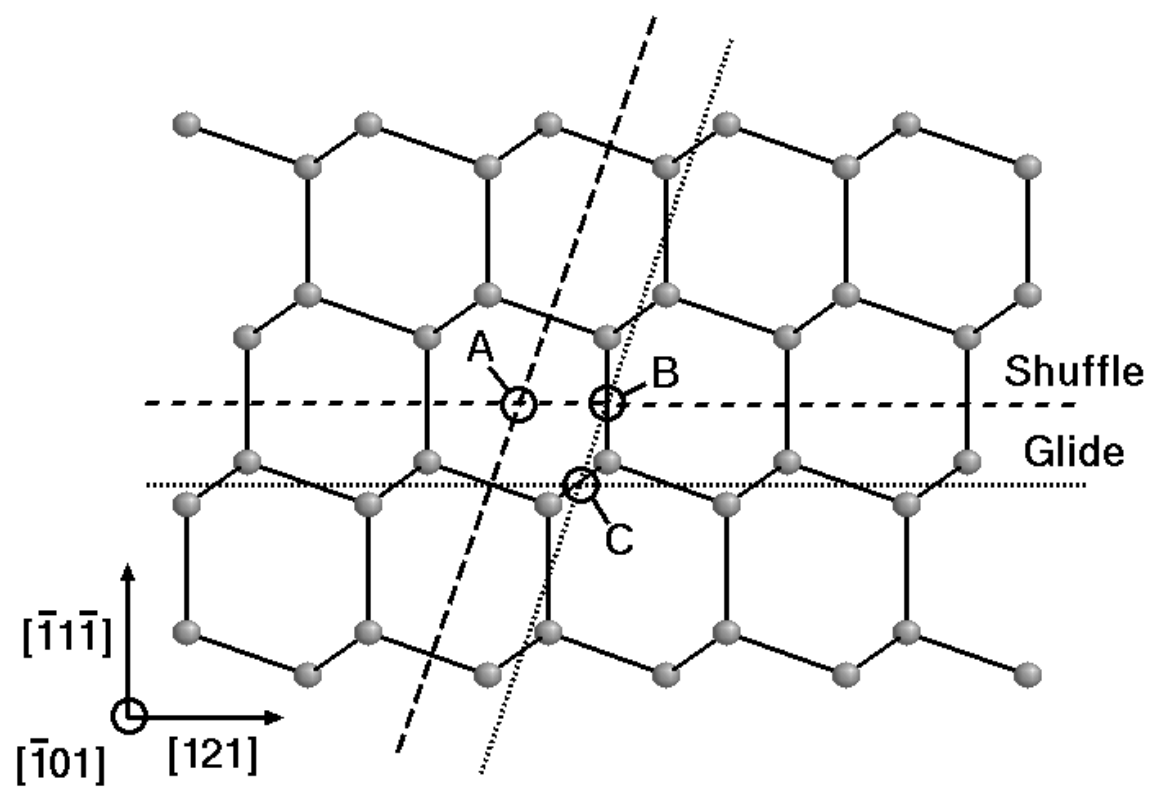


Figure 2

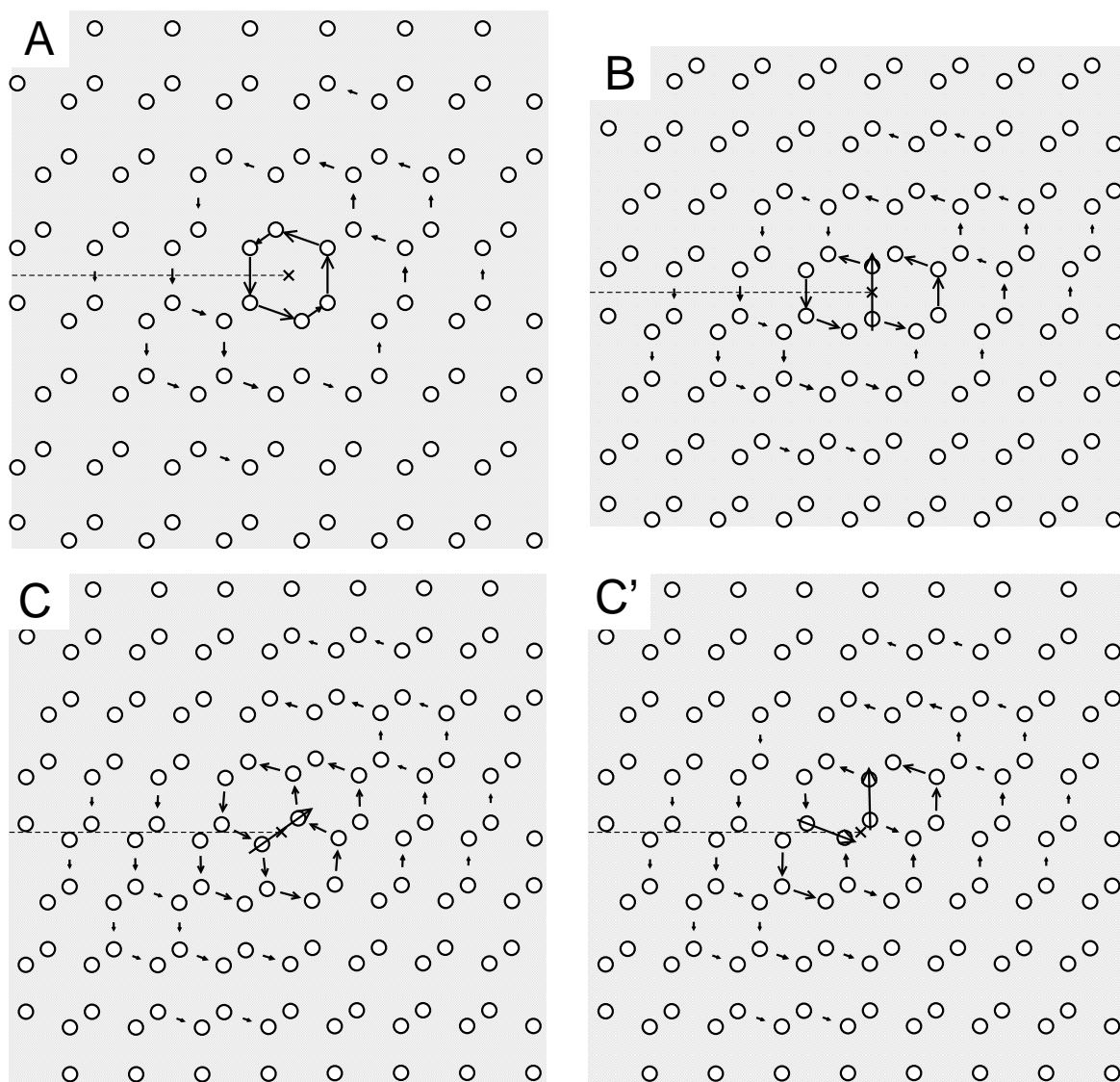


Figure 3

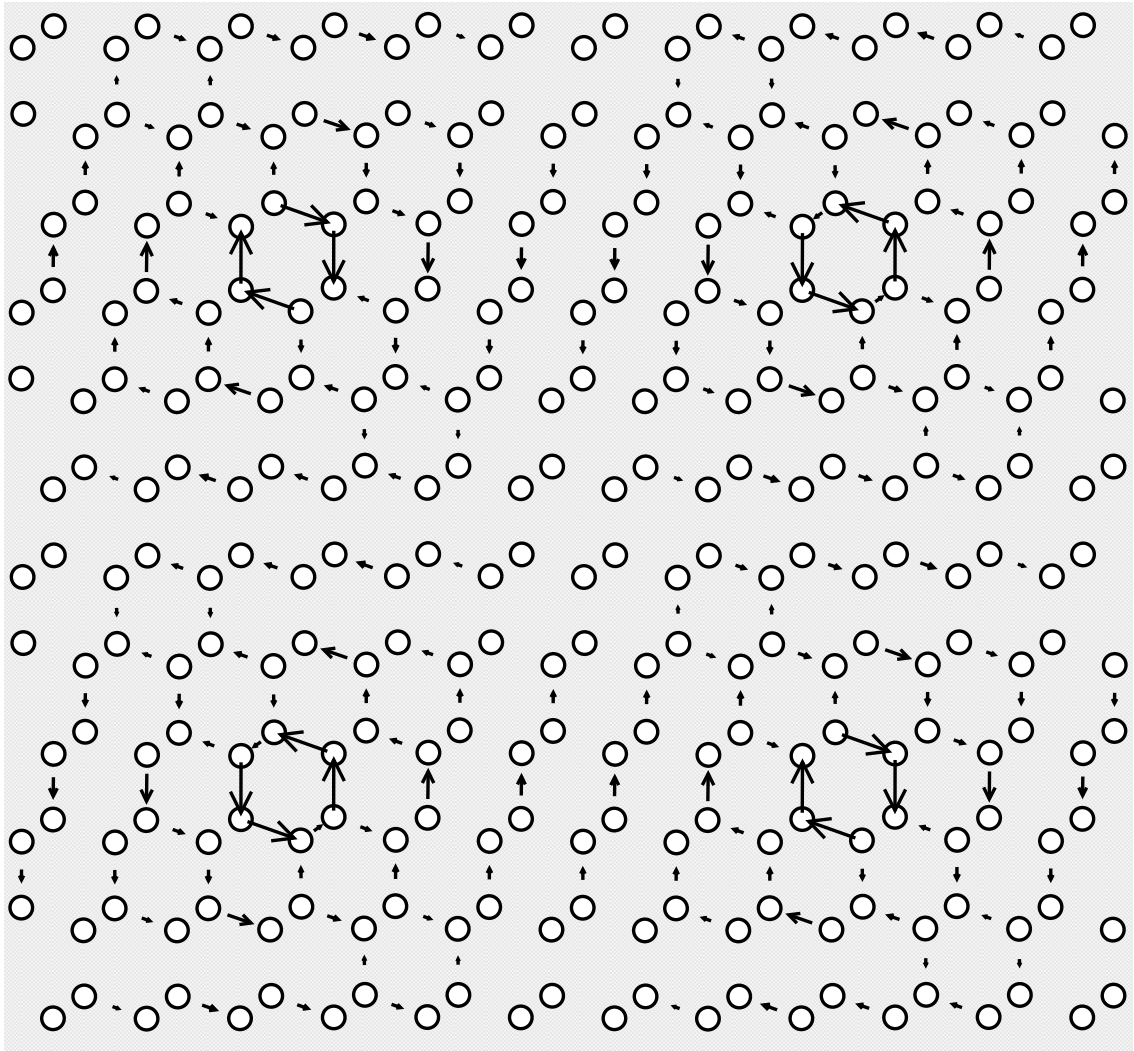


Figure 4

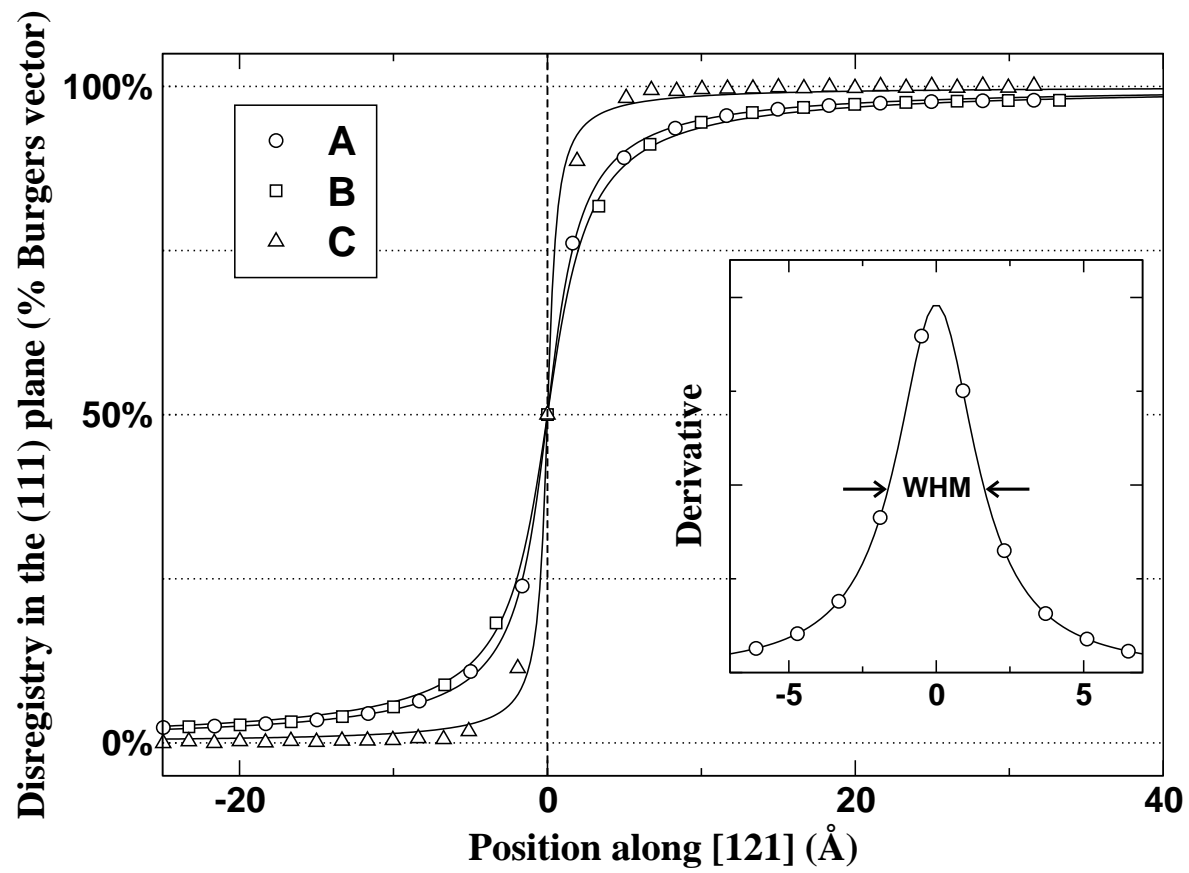


Figure 5

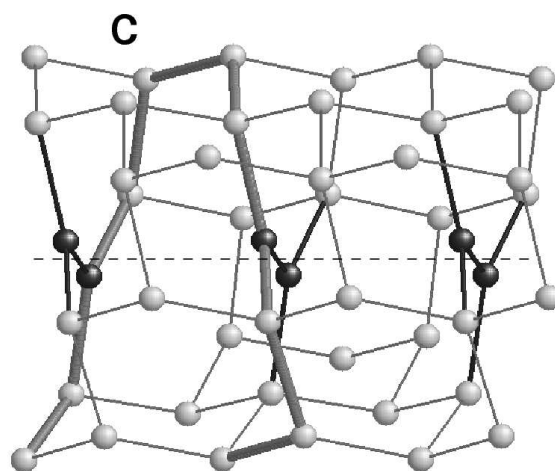
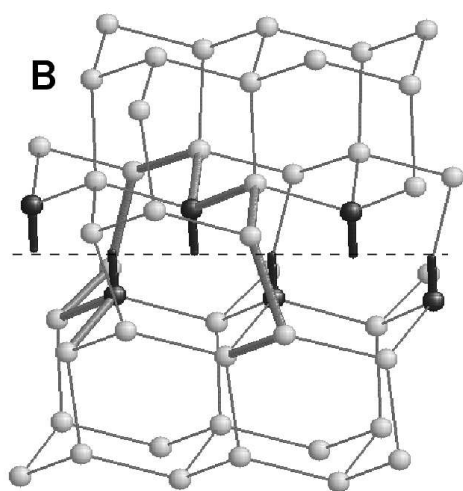
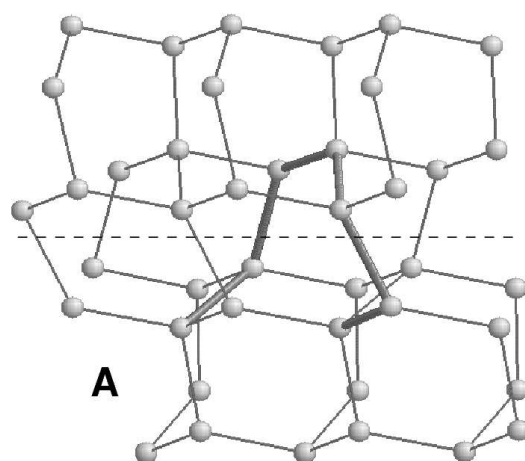
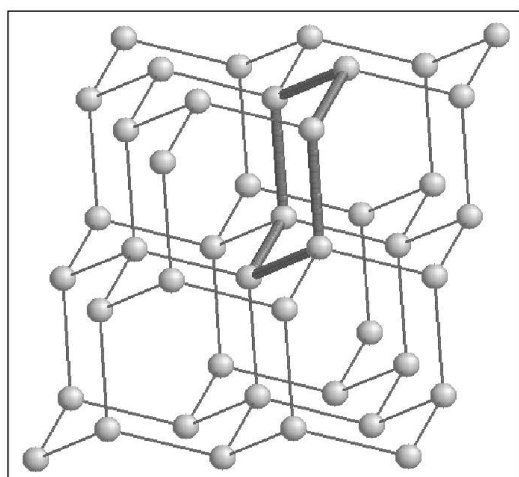


Figure 6